GEORGIA INSTITUTE OF TECHNOLOGY — OFFICE OF CONTRACT ADMINISTRATION

PROJECT ADMINISTRATION DATA SHEET

[X] ORIGINAL [ ] REVISION NO. ______

Project No. E-19-695 GTRI/~~XXX~~ DATE 6 / 16 /83

Project Director: Dr. Stephen D. Antolovich School/~~Lab~~ ChE

Sponsor: General Electric - Aircraft Engine Business Group, Evendale Plant

Type Agreement: P. O. No. 200-4-14C27795

Award Period: From 4/1/83 To 5/31/84 (Performance) ______ (Reports)

Sponsor Amount:

	This Change	Total to Date
Estimated:	$	$ 30,000
Funded:	$	$ 30,000* Fixed price

Cost Sharing Amount: $ ______ Cost Sharing No: ______

Title: Fractographic and Metallographic Analysis of INCONEL 718 Test Specimens

ADMINISTRATIVE DATA

OCA Contact Frank H. Huff

1) Sponsor Technical Contact:

2) Sponsor Admin/Contractual Matters:

K. R. Yockey, Buyer
Supplies and Contracts Procurement
Mail Drop A 182 N
General Electric Company
Aircraft Engine Business Group - Neumann Way, Building B
Cincinnati, Ohio 45215 (513) 243-2000

Defense Priority Rating: ______ Military Security Classification: ______
(or) Company/Industrial Proprietary: ______

RESTRICTIONS

See Attached ______ Supplemental Information Sheet for Additional Requirements.

Travel: Foreign travel must have prior approval – Contact OCA in each case. Domestic travel requires sponsor approval where total will exceed greater of $500 or 125% of approved proposal budget category.

Equipment: Title vests with Sponsor but none proposed.

COMMENTS:

*Only $15,000 can be billed in 1983.

This is a firm fixed price contract.

RECEIVED JUN 1983

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FORM OCA 4:383

GEORGIA INSTITUTE OF TECHNOLOGY OFFICE OF CONTRACT ADMINISTRATION

SPONSORED PROJECT TERMINATION/CLOSEOUT SHEET

5F990

Date 6/7/85

Project No. E-19-695 School/~~Lab~~ CHE

Includes Subproject No.(s) N/A

Project Director(s) S. D. Antolovich GTRC / ~~GIT~~

Sponsor General Electric-Aircraft Engine Business Group, Evendale Plant

Title Fractographic and Metallographic Analysis of INCONEL 718 Test Specimens

Effective Completion Date: 5/31/84 (Performance) 5/31/84 (Reports)

Grant/Contract Closeout Actions Remaining:

- [XXX] None
- [ ] Final Invoice or Final Fiscal Report -Already submitted.
- [ ] Closing Documents
- [ ] Final Report of Inventions
- [ ] Govt. Property Inventory & Related Certificate
- [ ] Classified Material Certificate
- [ ] Other

Continues Project No. Continued by Project No.

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FORM OCA 69.285

FRACTURE AND FATIGUE RESEARCH LABORATORY
Georgia Institute of Technology
A UNIT OF THE UNIVERSITY SYSTEM OF GEORGIA
ATLANTA, GEORGIA 30332
404/894- 2816

October 21, 1983

Dr. Tom Cook
GE Aircraft Engine Business Group
Building 200 G 60
Evendale, OH 45215

Dear Tom:

Enclosed is the report on the fractography of the IN 718 FCP specimens. We have recomputed the stress intensities using the data you supplied to us. To remain consistent we used the "old" formula for the stress intensity parameter. In any case, the main conclusions remain essentially unchanged. As we discussed, you have the original photos and I have not included any pictures in this report.

If you have any questions, just give a call. We're looking forward to seeing you in Atlanta.

Sincerely,

Stephen D. Antolovich, Director
Fracture and Fatigue Research Laboratory
Professor of Metallurgy

Enclosure

SDA/b

THE EFFECT OF TEST VARIABLES ON THE FRACTOGRAPHY OF IN 718 OVERLOAD FCP SPECIMENS

Prepared by

Eui W. Lee

and

Stephen D. Antolovich

Submitted to

T. Cook
General Electric Co.

November 1983

BACKGROUND

This report documents the results of a SEM fracture surface analysis study that was conducted on Inconel 718 specimens fatigued using various loading patterns. All specimens and mechanical test results were supplied by T. Cook, General Electric Co., Evendale, Ohio.

PROCEDURE

Fracture surfaces of samples subjected to various loading profiles were examined in a Cambridge SEM. Each specimen was first examined at low magnification (about 25X) to get an overview of the fracture and to see if there were any obvious differences. High magnification (approximately 1250X to 2500X) photographs were taken ahead of the starter notch in the depth (d) direction within an angle of ± 5 degrees to avoid stress intensity range variation due to the c direction component. The distance from the edge of the sample to the position where the pictures were taken was measured carefully using a micrometer attachment in the SEM. This was done so that an accurate stress intensity could be calculated for each position.

First, the fracture features formed at the same base line stress intensity range (ΔK_c) for each specimen were examined. Second, the visible transition region to tensile-dominated rupture associated both with the constant amplitude cycling, ΔK_b, and with the overload stress intensity range (ΔK_{OL}) were investigated.

RESULTS

For all the samples, fatigue striations were observed at $\Delta K \approx 25$ Ksi in$^{\frac{1}{2}}$. The striations were more distinct at low temperature than at high temperature and were more evident when the number of constant amplitude cycles between overloads (n) was 10 rather than 100 (Figs. 1 and 2). This is due to the fact that is a greater number of overload cycles when n = 10 compared to n = 100. The morphology of the fracture surface becomes increasingly noncrystallographic with increasing test temperature (Fig. 1a, 1b). For the specimens subjected to a longer hold time at maximum ΔK, cracks are increasingly associated with striations and secondary cracks are formed in the striations normal to the main fracture surface (Fig. 3).

On the fracture surface, distinct transitions from fatigue to final rupture were observed. Although the nature of transitions was the same, two different types of origin were identified. One is a complete transition associated with ΔK_b. The other is a local transition controlled by ΔK_{OL}. The complete transition ΔK_b is defined here as the value of ΔK above which the fracture surface is composed entirely of noncrystallographic overload rupture features. Local transitions associated with ΔK_{OL} are observed before the complete transition occurs. Table I shows ΔK_b at which complete transitions were observed for n = 100, for n = 10 and for samples subjected to a longer hold time at maximum ΔK. The effects of each variable (temperature, n, overload ratio, hold time at maximum ΔK) on the transition ΔK_b can be summarized as follows:

1. The transition occurred at marginally higher values of ΔK_b, the lower the temperature for a given overload ratio. This effect is more significant for ΔK_{OL}. Typical transition fracture surface features are shown in Figs. 4 and 5.

2. The transition occurred at higher ΔK_b when n = 100 than when n = 10 at the same temperature.

3. The transition ΔK_b decreased with increasing overload ratio. This effect was more significant for n = 10 compared to n = 100. Figures 4a, 4c, 5a and 5c show features associated with this effect.

4. A longer hold time significantly increased the transition ΔK_b and Fig. 6 shows the fracture surface features.

Table II shows ΔK_{OL} at which local transitions due to the overload were observed. The transition ΔK_{OL} due to overload was always much higher than for the complete transition ΔK_b. Figure 7 shows representative local transition fracture surface features in an area where a local transition was first observed. However, the effects of temperature, n and overload ratio on the transition ΔK_{OL} were all the same as the effects of those variables on the complete transition ΔK_b. For 50% overload and n=10, no local transition due to overload was observed. In this case, there was only a complete transition.

SUMMARY

The beneficial effects of overload usually found at room temperature seem to disappear at high temperature because the residual stresses induced by overload are apparently rapidly relieved at high temperature. One possible beneficial effect of overload at high temperature may be the increased fatigue crack closure since striations associated with oxide build-up and a rough fracture surface due to the overload may increase contact asperity. Effectively both oxide-induced closure and roughness-induced closure are operative at high temperature.

The transition is strongly dependent on the variables temperature, n, overload ratio, and hold time at maximum ΔK. It may be possible to use observations such as those cited in this report to ascertain the effects of various variables on high temperature fatigue life. However, a more complete statistical data base would be required.

TABLE I. Stress Intensity Range (ΔK_b) at Which the Fracture Surface is Tensile-Dominated

1. n = 100

	TEMPERATURE	
OVERLOAD RATIO	800°F	1200°F
50%	58.5 (4-81)	56.9 (4-82)
100%	-----	50.5 (4-83)
100%	52.9 (4-85)	51.4 (4-84)

2. n = 10

	TEMPERATURE	
OVERLOAD RATIO	800°F	1200°F
50%	54.6 (4-96)	53.3 (4-91)
100%	35.0 (5-88)	31.3 (4-97)

3. Long Hold Times at 1000°F

No overload (4-92) : 57.2 Ksi $in^{\frac{1}{2}}$
No overload (3-80) : 65.6 Ksi $in^{\frac{1}{2}}$
15% overload (6-99) : 58.0 Ksi $in^{\frac{1}{2}}$

TABLE II. Overload Stress Intensity Range (ΔK_{OL}) at Which Transition Occurs. At the Indicated Overload Values Tensile Features Are First Observed Locally.

1. n = 100

	TEMPERATURE	
OVERLOAD RATIO	800°F	1200°F
50%	82.6 (4-81)	68.2 (4-82)
100%	-----	57.5 (4-83)
100%	60.5 (4-85)	55.3 (4-84)

2. n = 10

	TEMPERATURE	
OVERLOAD RATIO	800°F	1200°F
100%	59.8 (5-88)	48.7 (4-99)

3. Long hold time at 1000°F

15% overload (6-99) ∵ 56.2 Ksi $in^{\frac{1}{2}}$

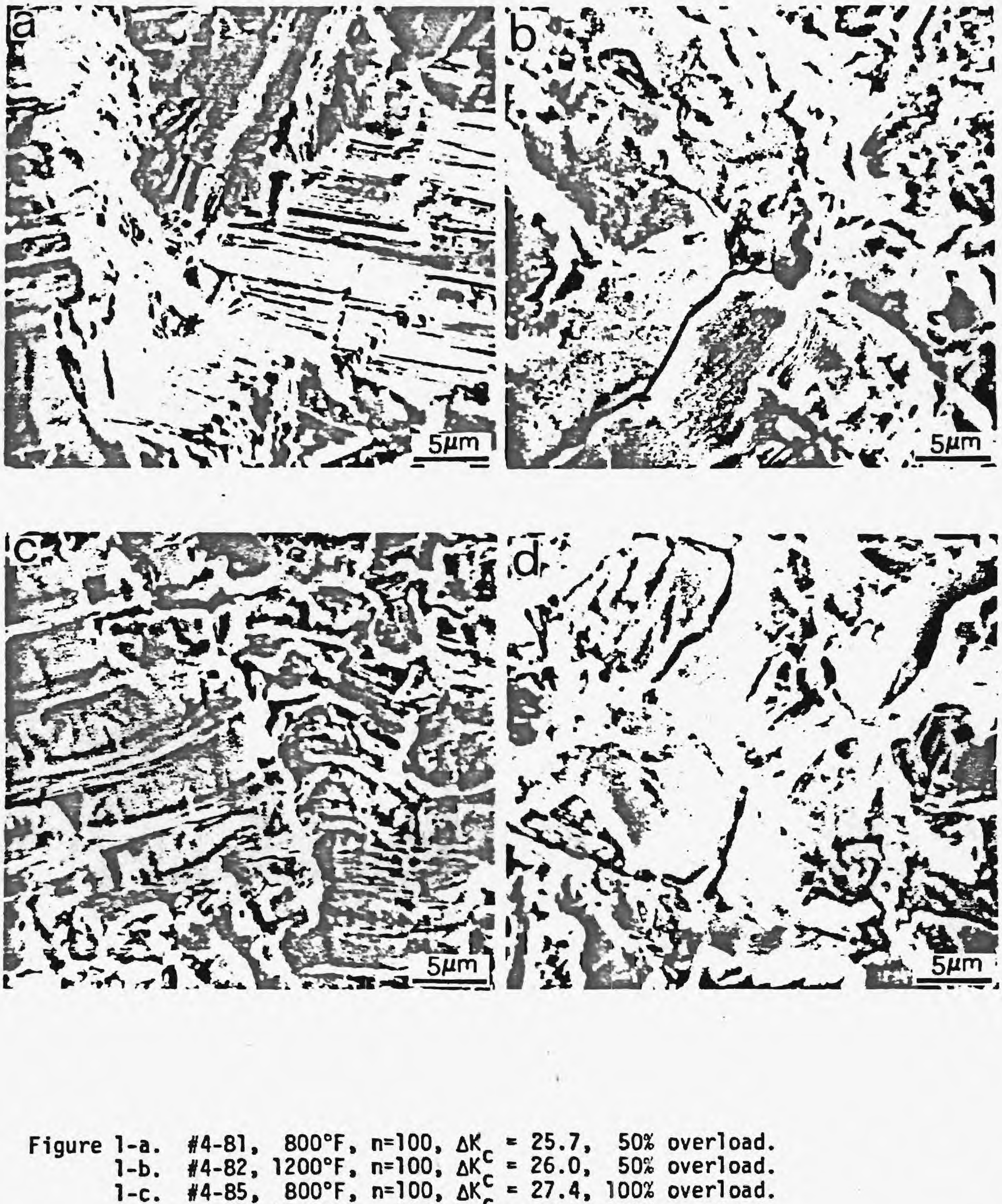

Figure 1-a. #4-81, 800°F, n=100, ΔK_c = 25.7, 50% overload.
1-b. #4-82, 1200°F, n=100, ΔK_c = 26.0, 50% overload.
1-c. #4-85, 800°F, n=100, ΔK_c = 27.4, 100% overload.
1-d. #4-84, 1200°F, n=100, ΔK_c = 26.5, 100% overload.

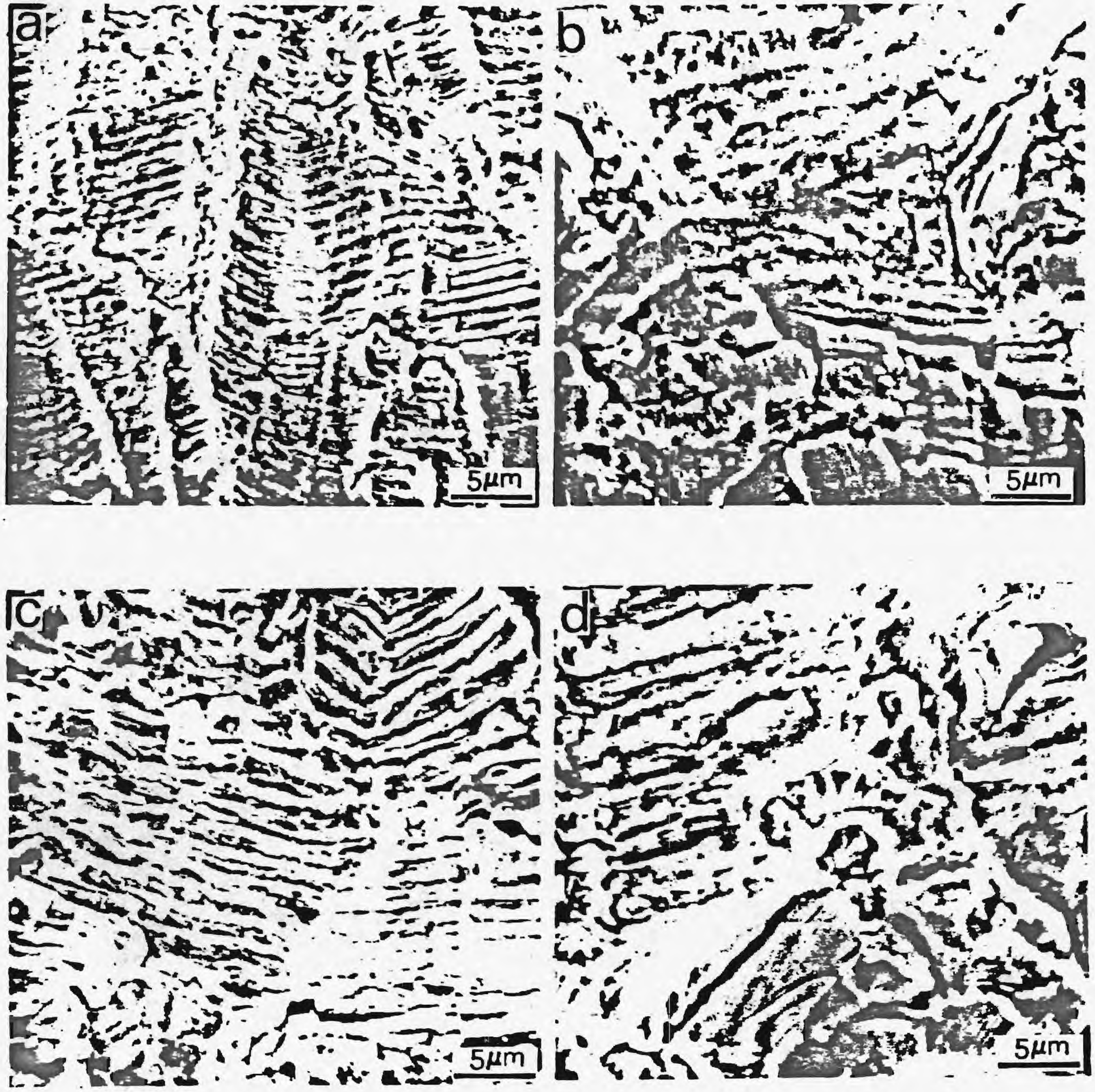

Figure 2-a. #4-96, 800°F, n=10, ΔK_c = 26.0, 50% overload.
2-b. #4-91, 1200°F, n=10, ΔK_c = 26.0, 50% overload.
2-c. #5-88, 800°F, n=10, ΔK_c = 26.5, 100% overload.
2-d. #4-97, 1200°F, n=10, ΔK_c = 27.1, 100% overload.

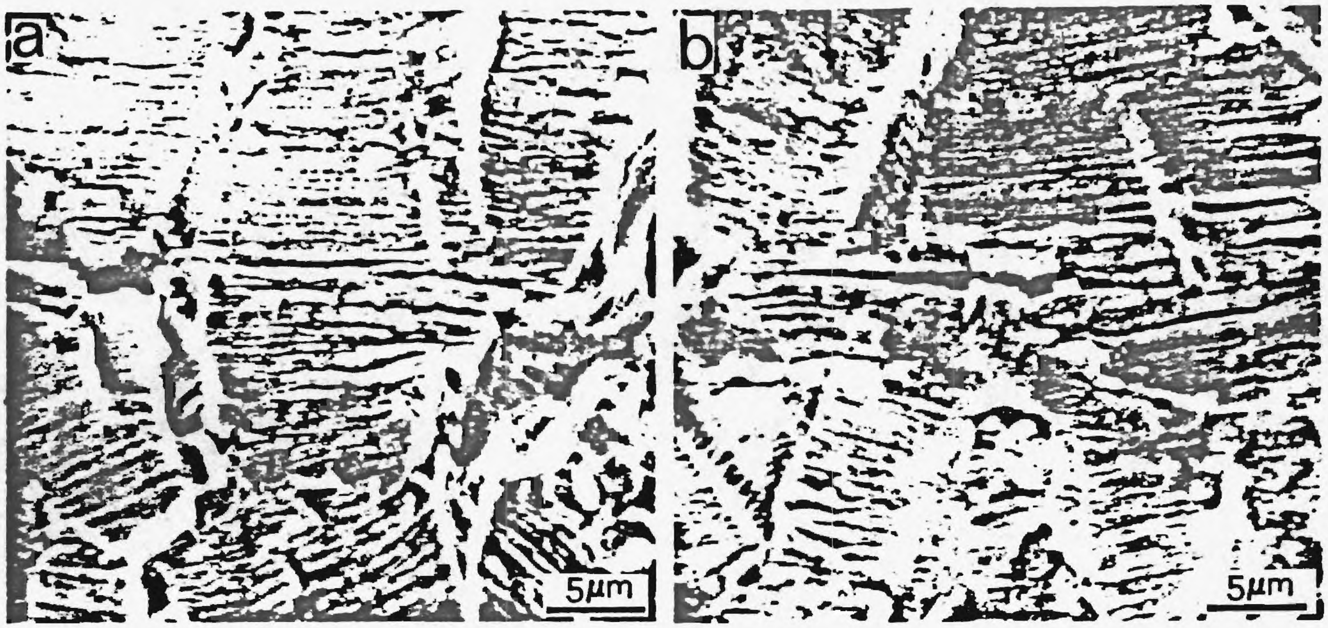

Figure 3-a. #3-80, 1000°F, 90 sec hold at max ΔK, $\Delta K_c = 25.2$.
3-b. #6-99 1000°F, 5 min hold at max ΔK, $\Delta K_c = 25.2$.

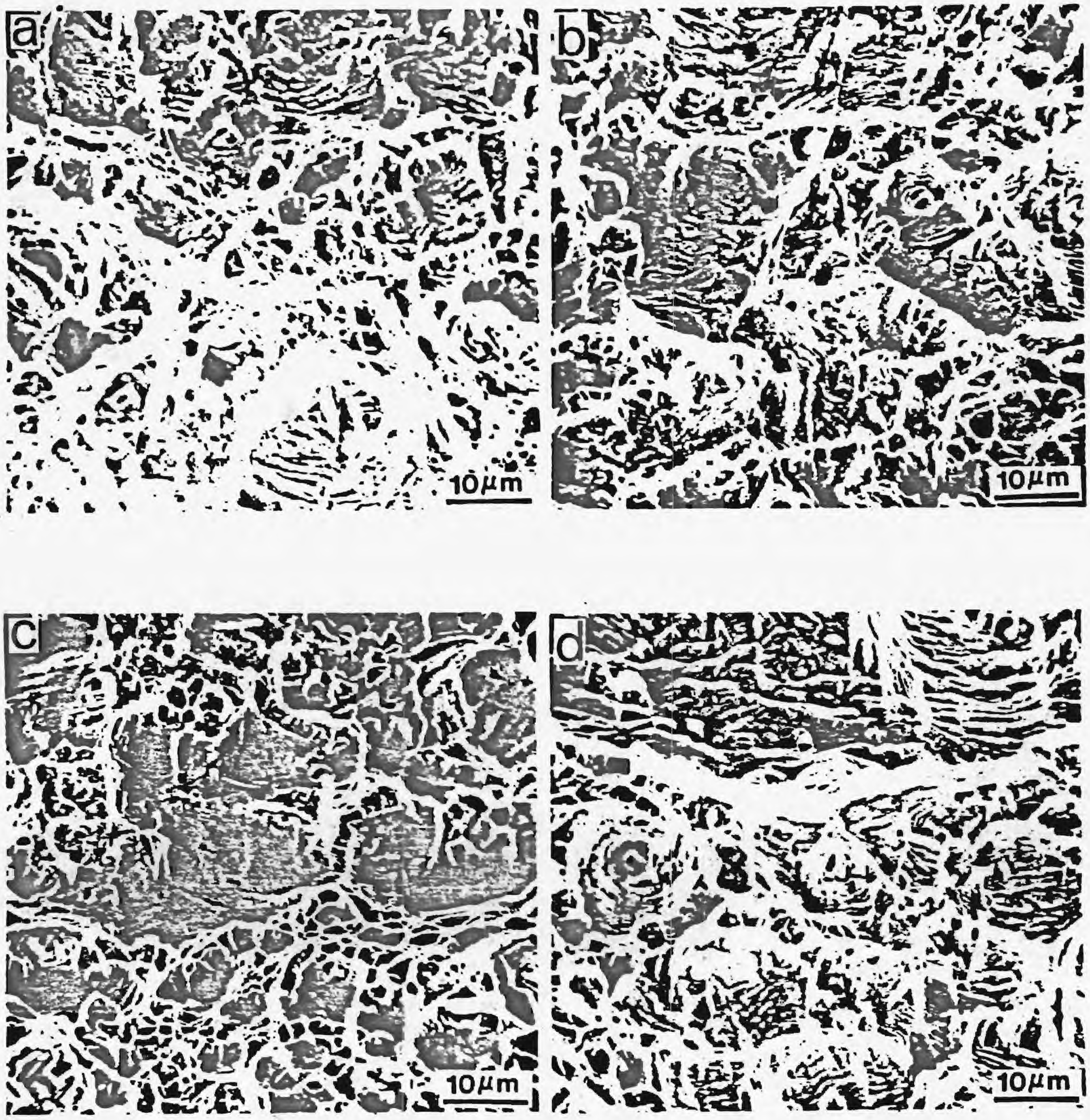

Figure 4-a. #4-81, 800°F, n=100, ΔK_b = 58.5, 50% overload.
4-b. #4-82, 1200°F, n=100, ΔK_b = 56.9, 50% overload.
4-c. #4-85, 800°F, n=100, ΔK_b = 52.9, 100% overload.
4-d. #4-84, 1200°F, n=100, ΔK_b = 51.4, 100% overload.

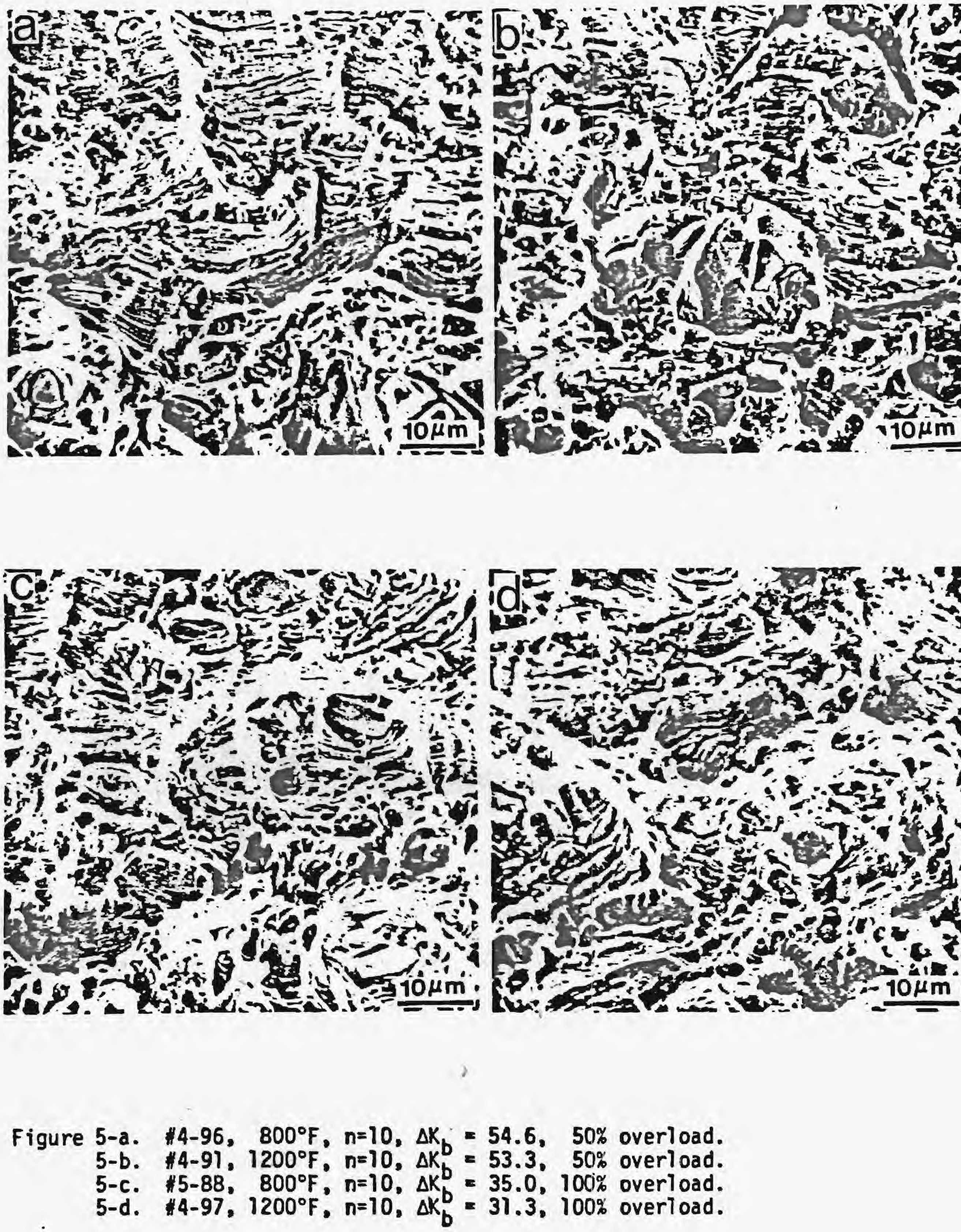

Figure 5-a. #4-96, 800°F, n=10, ΔK_b = 54.6, 50% overload.
5-b. #4-91, 1200°F, n=10, ΔK_b = 53.3, 50% overload.
5-c. #5-88, 800°F, n=10, ΔK_b = 35.0, 100% overload.
5-d. #4-97, 1200°F, n=10, ΔK_b = 31.3, 100% overload.

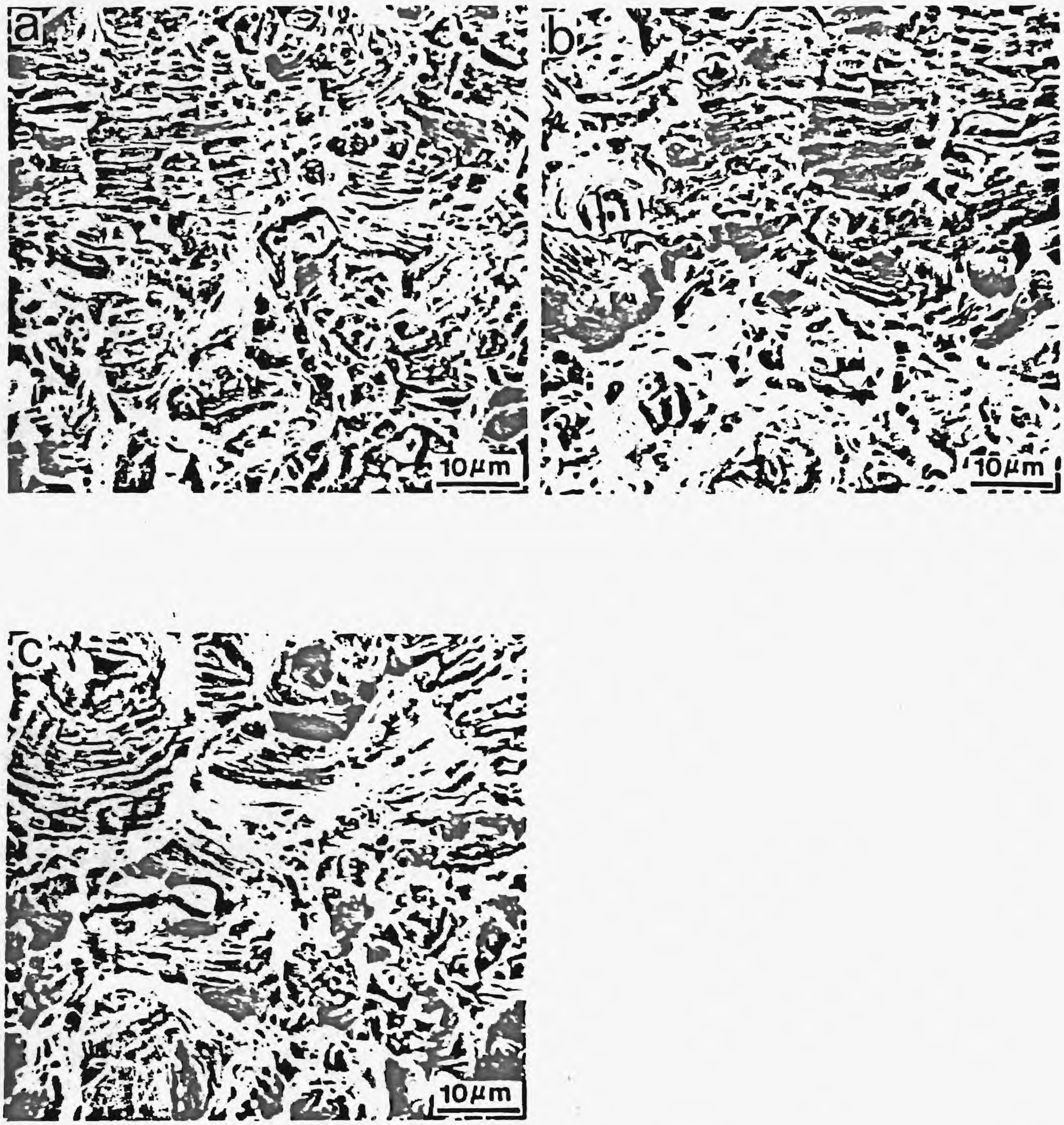

Figure 6-a. #3-80, 1000°F, 90 sec hold at max ΔK, ΔK_b = 65.6.
6-b. #6-99, 1000°F, 5 min hold at max ΔK, ΔK_b = 58.0.
6-c. #4-92, 1000°F, 5 min hold at max ΔK, ΔK_b = 57.5.

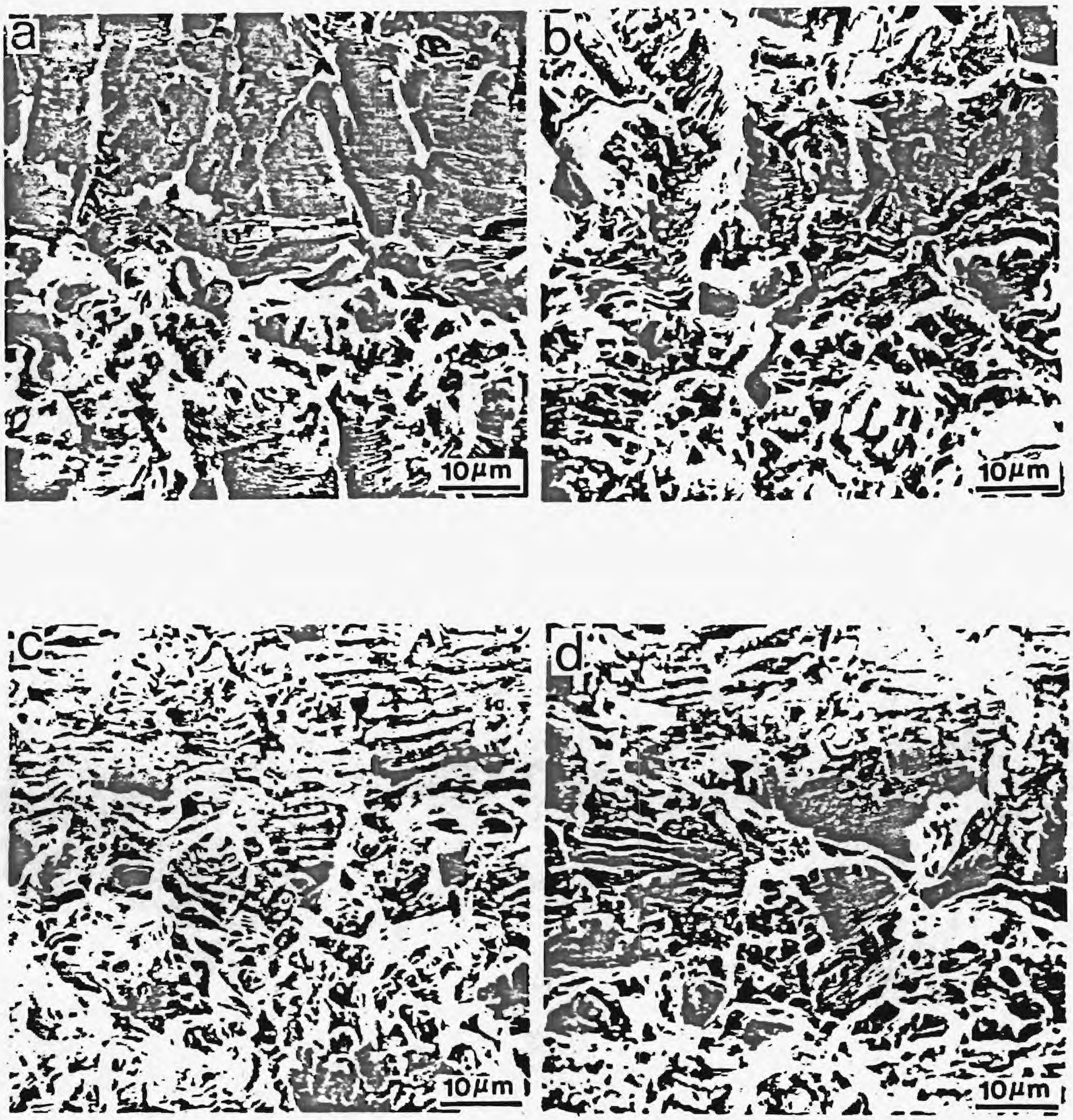

Figure 7-a. #4-81, 800°F, n=100, ΔK_{OL} = 82.6, 50% overload.
7-b. #4-82, 1200°F, n=100, ΔK_{OL} = 68.2, 50% overload.
7-c. #5-88, 800°F, n=100, ΔK_{OL} = 59.8, 100% overload.
7-d. #4-97, 1200°F, n=100, ΔK_{OL} = 48.7, 100% overload.

FRACTURE AND FATIGUE RESEARCH LABORATORY
Georgia Institute of Technology
A UNIT OF THE UNIVERSITY SYSTEM OF GEORGIA
ATLANTA, GEORGIA 30332
404/894-2816

November 28, 1984

Dr. Thomas S. Cook, Engineer
Materials Life and Methods
General Electric Company
Aircraft and Engine Group
Neumann Way
Cincinnati, Ohio 45215

Dear Tom:

I would like to review the status of the IN 718 project and let you know what our plans are for the upcoming year.

As you know, we are still awaiting machining of a large number of specimens from Low Stress Grind. After being machined, these specimens still have to be tested (as outlined in our original research plan) and examined via TEM. By necessity, we concentrated on the areas of aging, microstructural characterization and studies of the deformation substructures of specimens that were tested. The results of these studies were sent to you in the form of an interim report dated February, 1984. Although the time frame in which the studies were to be done has lapsed, I would like to assure you that as soon as we are supplied tested specimens by GE we will carry out metallurgical analysis and model development. in fact, as I indicated to you, we are willing to do some of the FCP testing at Georgia Tech to expedite completion of the project. I will support a student from my (limited) discretionary funds through the remainder of this academic year and will purchase all necessary supplies at no additional cost to GE. I feel that these studies along with those of Dan Krueger will form a good basis for understanding FCP in IN 718.

Thomas S. Cook, Engineer
Materials Life and Methods
General Electric Company
Cincinnati, Ohio 45215
November 28, 1984

You also supplied a number of IN 718 specimens to me that had been tested in fatigue at two different temperature levels. We chose a few specimens for microstructural analysis and the results are given in the attachment to this letter. The TEM work was carried out by Mr. Wally Milligan an M.S. student working on a NASA sponsored project. Mr. Milligan plans to do a small amount of additional analysis on the deformation substructures.

If you have any questions or would like more information, please do not hesitate to contact me.

Sincerely,

Steve

Stephen D. Antolovich, Director
Fracture and Fatigue Research Laboratory
Professor/Head of Metallurgy Program

SDA:cdj
Attachment

P.S.: Please check to see if GE has spare clevis type grips for the FCP specimens you sent to us.

AN INVESTIGATION OF NON-ISOTHERMAL FATIGUE IN IN 718

Walter W. Milligan
Fracture and Fatigue Research Laboratory
Georgia Institute of Technology

November 28, 1984

Introduction

It has become desirable to test gas turbine materials under non-isothermal LCF conditions in order to simulate actual engine conditions. Since a true Thermo-Mechanical Fatigue (TMF) test is both expensive and hard to interpret mechanistically, less complex testing is sometimes employed. In the present study, In 718 specimens were cycled isothermally at an intermediate temperature (1050 F) for a given fraction of the expected life, then unloaded, cooled to a low temperature (650 or 70 F) and cycled isothermally to failure. Several of these specimens were evaluated metallurgically to identify temperature or history-dependent damage mechanisms.

Experimental

The specimens which were evaluated are indicated in Table I, along with the relevant test parameters. The failed specimens were sectioned longitudinally, mounted, polished, and evaluated metallographically. Thin foils were prepared from each specimen and studied by Transmission Electron Microscopy. The microstructure is documented in Fig. 1.

Results

Data

As shown in Table I, cyclic lives for each of the specimens evaluated were within a factor of 2 (centered at about 7000 cycles).

As shown in Fig. 2, The data for 5/7 of the tests can be represented very well by a Coffin-Manson plot. On this plot, the high plastic strainrange and fraction of life spent at that plastic strain are plotted (if the plastic strains were significantly different between the two temperatures). This approach does not work in the last two cases, where the specimens were cycled at high plastic strainranges in the beginning of the test and low plastic strainranges at the end of the test.

There does not seem to be any effect of temperature, or relative time spent at the high temperature, on life. However, it would be useful to compare these tests with baseline isothermal data to substantiate this observation.

Metallography

No differences in behavior were observed metallographically. Failure was essentially transgranular, and oxidation damage was minimal.

TEM

Preliminary results indicate that the deformation mechanisms were independent of temperature. Slip bands were observed in all foils (Fig. 1). Further work on evaluating these foils is planned for December 1984.

Discussion

It can be concluded from the data and metallurgical analysis that the deformation and fatigue behavior of the alloy were independent of the change in temperature in the range that was studied. This is a natural consequence of the fact that similar deformation mechanisms (planar slip band formation and motion) were active at these temperatures. Since the environmental effect was small, the dislocation debris was the major source of fatigue damage. Although changes in this dislocation debris probably did occur between 70-1050 F, the magnitude of the change was relatively small, resulting in temperature-independent fatigue behavior.

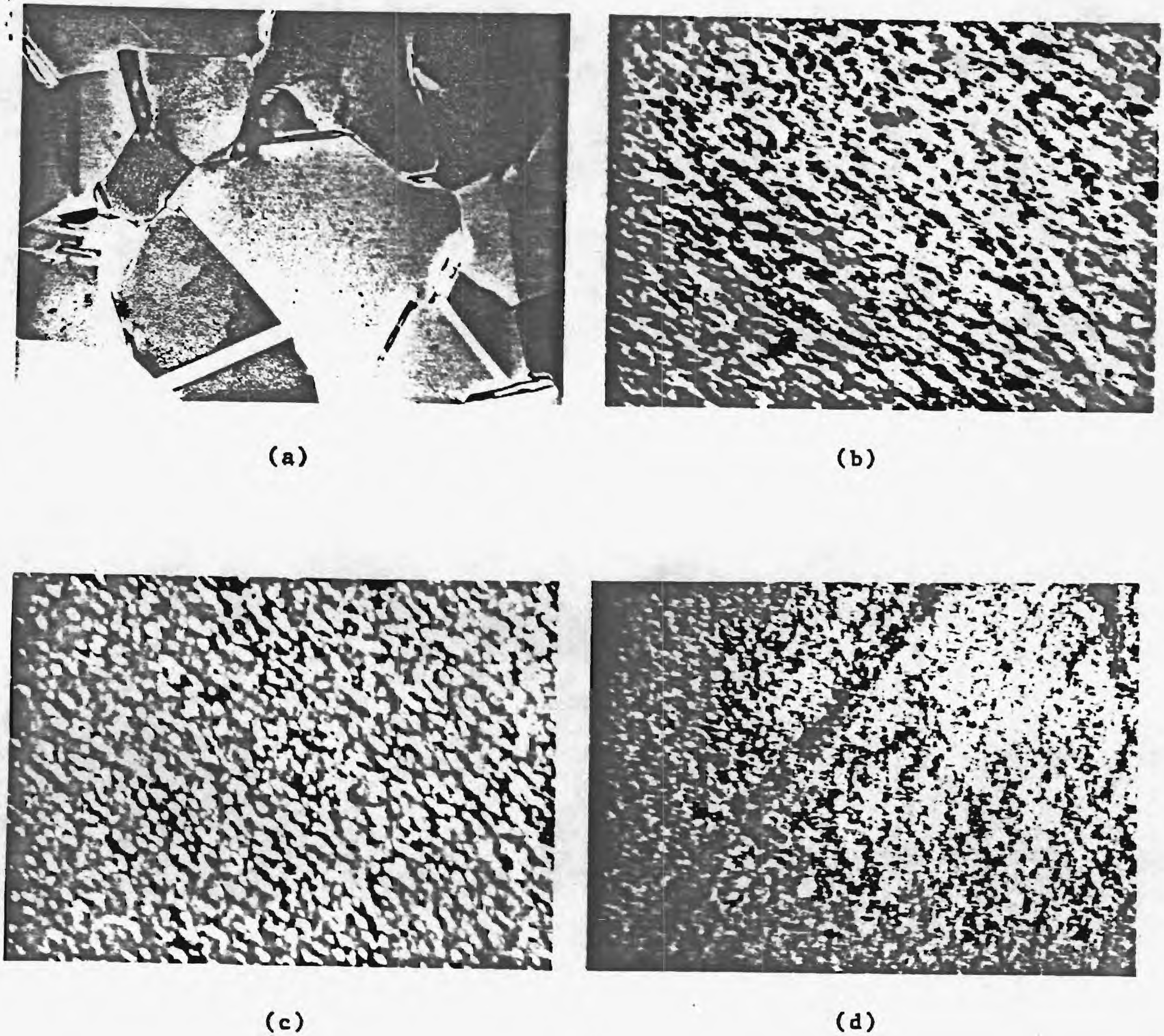

Figure 1. (a) TEM micrograph of typical grain structure, 5000X
(b) TEM micrograph of pre-tested microstructure, showing lattice strain contrast, 100,000X.
(c) Darkfield TEM micrograph of pre-tested microstructure. The γ'' was imaged by using the <300> reflection. 100,000X
(d) Darkfield TEM micrograph of the deformation microstructure, showing slip bands. The <300> γ'' reflection was used, 20,000X.

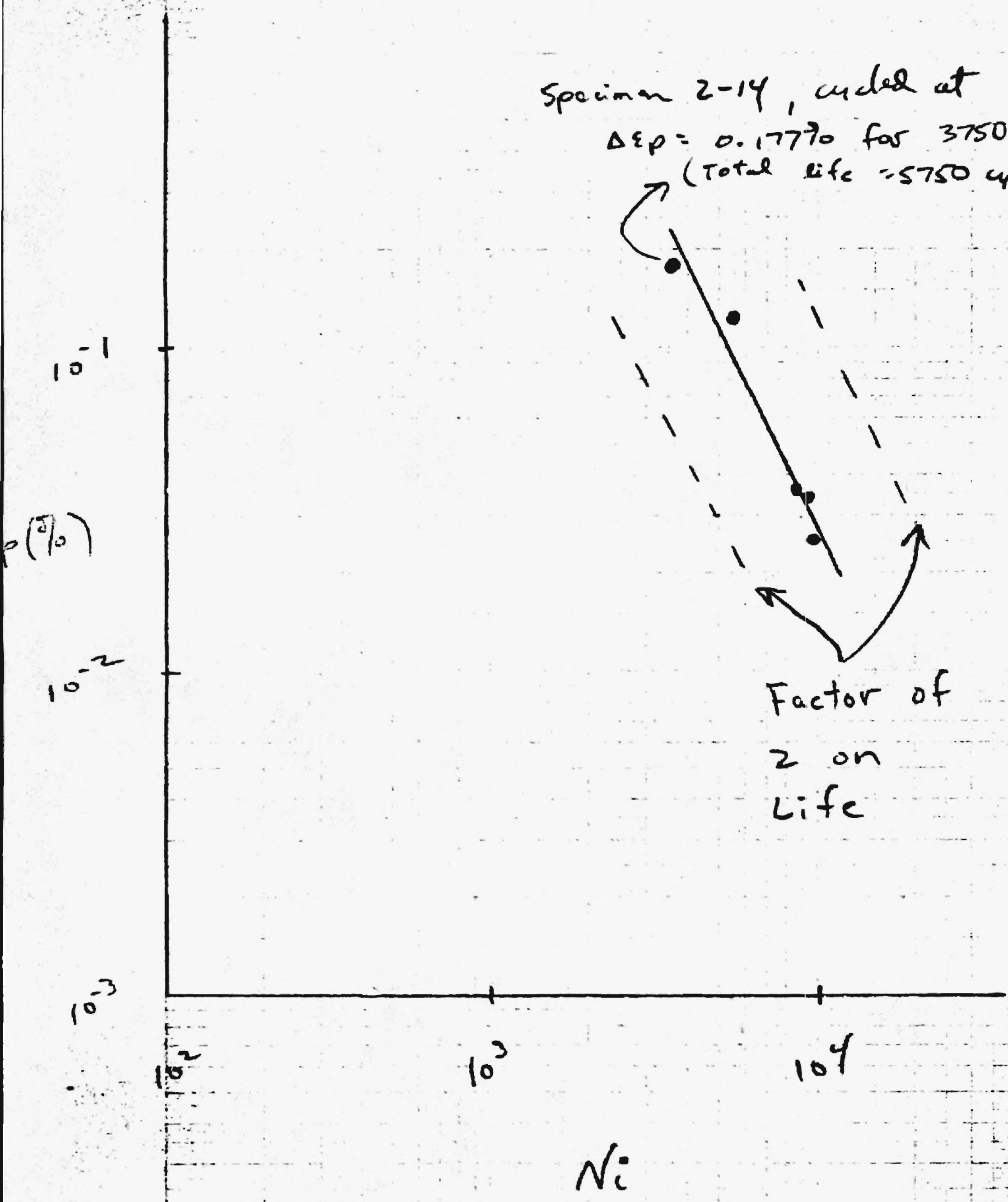

Fig. 2 - Coffin Manson Plot

TABLE I - FATIGUE DATA

SPECIMEN	T_1 (°F)	$\Delta\epsilon p$(%)	CYCLES	T_2(°F)	$\Delta\epsilon p$(%)	Ni
1- 9	1050	.035	100	70	.035	9300
1-10	1050	.090	100	70	.125	5500
2-15	1050	.035	3130	650	.040	8992
2- 2	1050	.025	5000	650	.025	9830
2-14	1050	.007	2000	650	.177	5750
1-19	1050	.150	1500	650	.018	4320
4-18	1050	.176	1000	650	.002	12,400
2- 4	1050	.035	5000	(Interrupted)		